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A residual stress study in similar and dissimilar welds

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Abstract

Residual strain distributions in similar and dissimilar welds were measured using neutron diffraction (ND) method. Then, using three strain components, three-dimensional stress states were calculated. The results were used to determine the effect of the martensitic phase transformation and material properties on residual stress (RS) distribution. It was observed that smaller longitudinal RS was induced in the low carbon steel side of dissimilar weld when compared to its similar weld. Also, it was found that the transverse RS near and within the weld zone (WZ) in dissimilar weld exhibited a distinctive trend, with tensile mode reaching the yield strength of the base metal (BM). In order to characterize the WZ in dissimilar weld, optical microscopy, hardness, energy dispersive X-ray spectroscopy (EDAX) were employed. This study not only provides further insight into the RS state in similar and dissimilar welds; it also delivers important consequences of phase transformation in the latter case.

Key word: Neutron diffraction, dissimilar weld, residuals strain and stress.

1. Introduction

Fabrication of structural components using dissimilar steels delivers structures that are lighter and more economical. Dissimilar weld joints such as stainless steel to carbon steel are being used in petrochemical, and power generation industries[1],[2].

Joining of dissimilar metals/alloys is generally more challenging than similar alloys[3],[4], which is due to several factors such as the differences in chemical compositions and physical properties. Furthermore, welding of dissimilar steels produces different residual stress (RS) distributions in welds as compared with welding of similar metals, and this merits investigation of RS distribution in dissimilar welds.

Several studies on the effect of RS on the failure of dissimilar weld joint have been reported[5]–[8]. For instance, Suzuki et al. [9] reported significant stress corrosion cracking as a result of RS in the dissimilar welds between ferritic steels and austenitic stainless steels, which are widely used in the oil and gas industry. Lately, a number of studies have used numerical models based on FEM analysis to predict RS in dissimilar welds. Deng et al.[10]determined the RS in a dissimilar metal pipe joint considering cladding, buttering, and post-weld heat treatment. Similarly, Lee et al.[11] predicted the axial and hoop RS produced in high strength carbon steel pipe weld using a FE model by employing a sequentially coupled 3-D thermal and solid-state phase transformation during welding. Generally, numerical techniques to estimate RS have been developed to a degree of sophistication, which are shown in aforementioned publications. However, material modeling has always been a critical issue in the simulation of welding because of the scarcity of material data at elevated temperatures. Some simplifications and approximations are usually introduced to cope with these problems. These simplifications are necessary due to both lack of data and numerical problems when trying to model the actual high-temperature behavior of the material. As a result, the effect of modeling idealizations can cause differences between modeling and experimental results. In addition, these discrepancies might increase when it comes to dissimilar welds because of uncertain material properties, uncertain chemical composition, and uncertain phase composition in the weld zone (WZ) and the heat affected zone (HAZ). In numerical modeling of dissimilar weld, these are normally ignored due to lack of the corresponding experimental data. For example, the European Network on Neutron Techniques Standardization for Structural Integrity (NET) showed the predicted stresses even in similar welds are sensitive to the numerical modeling assumptions and boundary conditions[12].

There are some previous works evaluating RS in dissimilar welds using experimental methods such as X-ray diffraction, hole-drilling, and ultrasonic measurement[13],[14]. However, neutron diffraction (ND) is outstanding among these techniques because of its ability to obtain RS non-destructively within the interior of components, in three dimensions, in small test volumes, and in thick specimens (up to several cm)[15]. In this study, in order to gain further insight into the RS of dissimilar welds, three welds (similar and dissimilar welds of low carbon steel and austenitic stainless steel) were created, and characterized using optical microscopy, hardness, energy dispersive X-ray spectroscopy (EDAX) and ND. This work is concentrated on RS distribution in dissimilar welds. It also explores how the material properties, phase transformation due to weld dilution in dissimilar weld can contribute in RS formation. Moreover, RS states in the corresponding similar welds were examined for comparison purposes.

2. Experimental

Material and welding procedures

Experimental work was carried out on three butt welds, using low carbon (AISI 1018) and austenitic stainless steel (AISI 304) plates. The experiment layout and material compositions of the alloys are shown in Tables 1 and 2, respectively. Single pass autogenous GTA welding was used for all cases. The welding parameters (displayed in Table 3) were designed so as to produce partial penetration since no auxiliary argon gas was accommodated for weld protection underneath. After welding, the specimens were allowed to cool for 2000 sec, after which time the temperature had approximately equilibrated, before being released from the C-clamps. Dimensions of samples are shown in Figure 1.

Microstructure of similar and dissimilar welds

It is important to analyze the microstructure of the welds prior to ND measurement in view of the fact that this measurement focuses on specific lattice plane spacings of a subset of grains, with specific orientations relative to the scattering vector (see Figure 2). This ability to actively select specific hkl lattice planes allows the separation of the strain responses of different phases in a multiphase material (Figure 2). In some cases, each phase in a multi-phase material, such as duplex steel, which roughly consists of a 50/50 mixture of austenite phase and δ -ferrite phase, can carry different stresses[15]. In this study, since the material consists of single phase, except the WZ of the dissimilar weld, only a single lattice plane is required to be used for the ND measurement.

Optical microscopy, micro-hardness measurements, EDS analysis, and the Schaeffer diagram were used to identify the phases compositions present in the weldment in similar welds and dissimilar welds. Since mostly ferrite and austenite were present in 1018 similar welds and 304 similar welds, respectively, a single lattice plane was used for each similar weld. The (211) and (311) planes are suitable for ferritic and austenitic plates since these two planes approximate the elastic strain response of the bulk material, and are not sensitive to intergranular strains [15]. In the ferritic-austenitic dissimilar weld, a new phase (martensite) was present in the WZ, the ND setup was not changed to capture this small region. Therefore, the measurement was run again with (211) plane in 1018 side and (311) plane in 304 side.

Neutron diffraction (ND) measurement

ND has benefits such as deep penetration (centimeters) into many engineering materials, three-dimensional mapping, and volume-averaged bulk measurement capabilities. These characteristics make ND a powerful tool for the measurements of RS at depth in welds[15]. In this study, the ND measurements were carried at the High Flux Isotope Reactor of Oak Ridge National Laboratory (HFIR) on the 2nd Generation Neutron Residual Stress Facility (NRSF2). The beam geometry and the samples mounted in the fixture are shown in Figures 3 and 4, respectively. The sample translation stage and goniometer are able to perform precision motions (0.01mm) for the placement of an instrumental gauge volume. The gauge volume is defined by a set of Gd slits. In the experiment, two different gauge volume geometries were utilized (Figure 3). The gauge volume geometry was chosen dependent on the

strain direction being measured. For the transverse and normal directions a gauge volume of 1mm x 1mm x 4mm (width, depth, and height) is used. The extended height is along direction of the weld and is a direction of symmetry in the sample. For the longitudinal case the gauge volume was changed to 1mm x 1mm x 1mm so the height would be coincident with the welding direction.

The incident wavelength for the instrument is defined by a bent silicon crystal focusing monochromator[16]. For the experiment a wavelength of 1.73 Å is used. This defines the Fe (211) reflection at a 2θ of ~95.1° for the 1018 side and the Fe (311) reflection at a 2θ of ~106° for the 304 side.

At each measurement point, lattice strains were calculated from the corresponding d-spacing changes in three orthogonal directions (shown in Figure 3) such that the stress fields could be calculated. The lattice strain can be determined from the measured lattice spacing according to [15]:

$$\varepsilon_{ii} = \frac{d - d_o}{d_o} = -\cot\theta(\theta - \theta_o) \quad i = 1,2,3 \quad \text{Equation 1}$$

where the d and d_o are the interplanar spacing under the stressed and stress-free state, respectively. And θ and θ_o are the diffraction angles for the stressed and stress-free specimens at each location, respectively[15]. Since d_o is required as a reference in ND strain calculation, it was determined using macroscopic “stress-free” comb-like coupon, as shown in Figure 4. Each comb-like coupon was cut from a nominally identical weld at the same locations as the ND measurements using electric discharge machining (EDM). The coupon has been sliced with 3 mm wide and 3 mm thick along the transverse direction in order to release the macroscopic stresses effectively from the bulk of the weldment. With the assumption that the three orthogonal components of measured strain correspond to principal directions, meaning the shear stresses are assumed zero in these defined directions, the analysis to determine RS is simplified. The macroscopic stress components, where the numerical superscripts refer to the lattice plane family, are related to the elastic strains in analogy with Hooke’s Law by[15]:

$$\sigma_{ij} = \frac{E^{hkl}}{(1 + \nu^{hkl})} \left[\varepsilon_{ij} + \delta_{ij} \frac{\nu^{hkl}}{(1 - 2\nu^{hkl})} \varepsilon_{kk}^{hkl} \right] \quad \text{Equation 2}$$

where E_{hkl} and ν_{hkl} are the diffraction elastic constants’ relating strain in the (311) and (211) lattice planes to the macroscopic stress. And also ε_{kk}^{hkl} and ε_{kk}^{hkl} in the above equations are the elastic bulk strains. These elastic constants may be either measured or calculated with a good accuracy based on the self-consistent Kröner model and using single crystal elastic constants for austenitic stainless steel and low carbon steel¹⁰. The values used in this study are shown in Table 4.

3. Results

3.1. Macro and Micro-structure studies of the welded samples

Macrostructure and microstructure of the three cases (described in Table 1), are shown in Figures 5, 6, and 7. For these three joints, it can be seen that the WZs are different from one another. For instance, the depth of penetration in similar 1018, similar 304, and 1018-304 dissimilar, are 1.7, 0.9, and 1.53mm, respectively. Different depths and weld shapes are attributed to the different thermal physical

properties of the BMs. In the dissimilar weld, the WZ and HAZ are asymmetric, and the deepest penetration is located off the weld centerline with larger WZ in 304 side as the thermal conductivity of the 304 is much less than 1018. This phenomenon is studied in detail by Bahrami et al.[17].

The Base metal (BM) of 1018 weldment has microstructures of ferrite and pearlite, as shown in Figure 5. The WZ contains coarse pearlite as results of intermediate cooling rate to room temperature, while HAZ contains fine pearlites due to faster cooling rate. Since ferrite was mostly present in the BM, HAZ, and WZ of 1018 similar weld, the (211) Bragg reflection was used for ND measurement.

Figure 6 shows the microstructure of the 304 weldment. It is well known that the microstructure of 304 is mainly composed of single phase austenite. However, this austenitic microstructure contains a small amount of δ -ferrite. Figure 6 also shows a columnar dendritic structure within the WZ. This is attributed to the fact that, during weld metal solidification, some supercooling effects occur. Since the austenite phase was available in all three regions (BM, HAZ, and WZ), the ND measurements used the (311) Bragg reflection.

Figure 7 shows the macro and microstructure of the 304-1018 dissimilar weld. The Schaeffler diagram was used to predict the microstructure of the WZ for dissimilar weld. To calculate dilution, cross sectional areas of the melted 304 BM and 1018 BM were calculated with the aid of ImageJ (image analysis software)[18]. Calculation showed approximately 43% dilution from 1018 and 56% dilution from 304 BM. Considering the composition of the BM, Cr_{eq} and Ni_{eq} were evaluated. Three points, representing the microstructures for the 1018, 304 and dilution of dissimilar weld, are shown in Figure 8. According to EDAX analysis, the result of which is shown in Table 2, Cr_{eq} and Ni_{eq} were calculated. These results are plotted in Figure 8, which are in good agreement with the measured dilution. According to Schaeffler diagram, dissimilar WZ is martensitic. Strain and stress carried by this phase was not captured with ND measurement since only ferrite and austenite phases were considered.

3.2. Micro-hardness test

Hardness profiles can assist the interpretation of weld microstructures and mechanical properties. In this work, several metallographic specimens were prepared from each weldment. Micro-hardness tests were performed by measuring values across the weld cross-section, crossing both HAZ and WZ for the three cases (Table 1). The Vickers test was measured with a load of 500 gf and a loading time of 10 s along both paths 1&2, as shown in Figure 9. Microhardness profiles across the three weldments are shown in Figure 10. From the profiles of Figure 10A (path1) and Figure 10D (path2), it was observed the hardness of similar 1018 weldment slightly increases from BM to WZ. This is due to increasing the density of pearlite within WZ. Hardness in the 304 similar weld, shown in Figure 10B (path1) and Figure 10E (path2), has essentially no variations throughout the weldment. However, in the case of the 304-1018 dissimilar weld a sharp increase (about 200HV) can be seen crossing the HAZ and WZ of the weldment. The reason behind this increase could be due to formation of martensite in the WZ, which confirms the microstructure determined by the Schaeffler diagram.

3.3. Neutron diffraction

3.3.1. Interplanar spacing (d) and stress-free spacing (d_0)

Figures 11 to 13 show the variation of the interplanar spacing (d) in three directions for similar and dissimilar welds measured by ND along mid-thickness of plates. Note that the interface between all three welds was marked as 0 mm. The results of the macroscopic stress-free, d_0 , which was measured from the reference coupon specimens, are shown within ± 25 mm away from weld centerline. The reason is that the chemical composition of base metal (BM) outside of this zone (± 25 mm), is far from the process affected zone, and d_0 does not change considerably. Therefore, the measured d_0 at location 25mm can be extended for the rest of the BM for strain and stress calculations. In Figures 11 to 13, the d of the longitudinal component is clearly higher than the other two components (transverse and normal). The asymmetry observed in the 304 similar weld d_0 measurement is most likely due to some heterogeneous gauge volume effects. This is expected in the 304 case due to heterogeneous changes in the phase fraction of δ -ferrite in the weld region. Typically if one extent of the gauge volume has a reduced amount of grains contributing to diffraction it will lead to an artificial shift of the gauge volume. In this case, the shift is due to an increase in δ -ferrite.

The d_0 values for the dissimilar weld are difficult to obtain and index versus the measured d -spacings. This is because the unstressed spacings are measured from an analog plate, where the 304-1018 transition is difficult to position, and may not occur the same position as in the dissimilar measurement sample. To overcome this challenge, a stress balance approach was used to define the actual center of the d_0 sample. The measured d_0 values and the interpreted values are shown in Figure 14.

3.3.2. Strain

Using Eq(1), three strain components (longitudinal, transverse, and normal) of the similar and dissimilar weldments are calculated and plotted in Figures 15 through 17, respectively. First, three strain components of two similar welds are addressed. Then, the results of dissimilar weld will be presented.

As anticipated, the longitudinal residual strain of all three cases is in tensile mode near the WZ (± 25 mm). The longitudinal strain width of the 304 weld is wider than of 1018 weld. This is mainly due to higher CTE of 304 than that of 1018. The transverse residual strains of both 1018 similar and 304 similar welds are in tensile mode, as well. However, their magnitudes are smaller than the longitudinal component. The reason is that the contraction due to material shrinkage in longitudinal direction is greater than that of transverse direction. The normal strain components of all three cases are in compressive mode (± 30 mm), due to the Poisson effect.

It is interesting to note that the longitudinal residual strain in 1018 side of dissimilar weld (Figure 15) experienced a residual strain reduction (between 10 and 20mm away from the weld center line). This fact was discussed by Eisazadeh et al.¹⁴. Another interesting point is that the transverse residual strain in dissimilar underwent a sharp increase near WZ (± 30). These two points will be discussed in more detail in the discussion section.

3.3.3. Stress

Using Eq(2), three stress components (longitudinal, transverse, and normal) of similar and dissimilar welds are calculated and plotted in Figures 17 through 19, respectively.

The individual stress components do not follow the same trends seen in the corresponding strain components, due to the contribution of three strain components in Eq(2). The aforementioned fact about stress reduction in dissimilar weld is clearly shown in Figure 18.

4. Discussion

4.1. Similar weld

The similar weld of 1018 steel shows that the maximum value of residual longitudinal stress is limited to a small region (± 7 mm) near the weld centerline (Figure 18). However, in the 304 similar weld this zone is extended to a wider area (± 20 mm). The reasons behind this fact are that 304 steel has a larger CTE and yield strength, yet lower thermal conductivity than the 1018 steel. Since the CTE has a significant effect on the RS, and this property of 304 is much greater when compared to the 1018 (see Table 5), the RS in the 304 weld is increased remarkably to a high value. Lower thermal conductivity of 304 could boost up the RS in the 304 sample since it leads to higher thermal gradients. Also, heat capacity is lower for 304 leading to higher temperatures for a given energy input. These cause the area near the weld line of the 304 sample to remain at high temperatures whereas away from weld line it remains cool[19]. Therefore, the region near the weld line experiences a high temperature gradient, which causes higher RS.

4.2. Dissimilar weld

Generally, when a dissimilar weld is carried out, residual stresses induced by the arc welding process could produce different RS distributions especially near the WZ. This is due to the different CTE and yield strength of the base metals, as described clearly in the previous study[20]. Figure 18 shows a large stress reduction on the 1018 side of dissimilar weld. This is because during the weld thermal cycle, the 304 plate, which has a larger CTE, can produce a tensile load on the 1018 side of the weld, while placing itself into a state of compression. It has been shown that the application of tensioning load during welding process can greatly reduce the longitudinal tensile RS in FSW joints[20]–[22].

Besides these two material properties, the microstructure of the WZ in a dissimilar weld can contribute significantly to RS formation near the weld interface[23]. In Figure 7 it was shown that the microstructure of the WZ in the dissimilar weld sample is mainly martensitic because of the varied composition and rapid cooling. This hypothesis was confirmed with the micro-hardness test, shown in Figures 10C and 10F. When the FCC structure of the WZ (austenite) transforms to the BCT structure (martensite) during cooling, a volume expansion is experienced. This increase produces a compressive stress within the WZ, however it induces a tensile stress to the surrounding area.

It was shown that the martensite was not present throughout the plate because of partial penetration of WZ, as shown in Figure 7. So, during ND measurement, the RS within the ferritic and austenitic phases of surrounding area were captured because the gauge volume overlaps the BM and the WZ. Although gauge volume includes the WZ (martensite) partially, the compressive stress within the WZ was not recorded because it was neither of these two the ferritic and austenitic phases and the ND measurement indicates only stresses within the selected phase within the gauge volume. This is shown in Figure 2. A single phase measurement was all that was carried out in this experiment. A complete representation of multiple phases' contribution to residual stress could be determined using a time-of-flight neutron diffractometer, such as the VULCAN diffractometer at the Spallation Neutron Source, and performing a Rietveld refinement to determine the lattice parameter shift, taking into account multiple phases[24],[25]. However, this method was not used in this study because the required neutron gauge volumes are currently unachievable at VULCAN.

5. Conclusions

In this study, residual strain and stress distributions of similar and dissimilar welds were measured using neutron diffraction. Effects of phase transformation and material properties on residual stress distributions in the dissimilar welds were addressed. The following conclusions are reached:

1. The tensile region of residual stresses in the longitudinal direction extends much further from the weld center in the case of the similar 304 weld than in the similar 1018 weld, due to the differences in thermal and mechanical properties between the two materials such as CTE, thermal conductivity and heat capacity.
2. Asymmetric residual stress distributions were observed in the dissimilar weld, which was caused by differences in CTE, yield strengths, and thermal conductivity of the two base metals.
3. The transverse residual stresses induced by welding exhibited distinctive trends between the similar and dissimilar welds, which is attributed to existence of martensite phase in the latter.

4. In the dissimilar weld, the thermal expansion coefficient plays a role on stress reduction in 1018 side by introducing tensioning load caused by the 304 plate.

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Table 1. Experiment layout

Cases	Material
Case-1	Similar weld of 1018 steel
Case-2	Similar weld of 304 stainless steel
Case-3	Dissimilar weld of 304 stainless steel & 1018 steel

Table 2. Chemical composition of steel plates (weight %)

Composition	C	Mn	Si	P	S	Cr	Ni	Balance	Ni _{eq}	Cr _{eq}
1018 steel (average)	0.13-0.20	0.30-0.90	0.15-0.30	0-0.04	0-0.05	-	-	Fe	≈6	0.3
304 stainless (average)	0-0.08	1.7	0.52	0-0.045	0-0.035	18.9	7.5	Fe	≈10	≈20
1018-304 WZ (EDAX)	-	1.14	0.12	-	-	7.58	4.32	Fe	≈7	≈8

Table 3. Welding parameters used in this study

Arc voltage (V)	Arc current (A)	Electrode diameter(mm)	Arc length(mm)	Travel speed (mm/s)	Argon gas (m ³ /s)	Arc efficiency
18	150	2.4	1.5	4	0.0136	80%[19]

Table 4. Diffraction elastic constants used in this study [15]

Material	<i>hkl</i>	E _{hkl}	ν _{hkl}
1018 steel	Fe (211)	225.5 GPa	0.28
304 stainless steel	Fe (311)	183.5 GPa	0.31

Table 5. Mechanical and thermal properties of 1018 and 304 at room temperature [21]

Material	Thermal conductivity, W/m. °K	Thermal expansion coefficient, 1x10 ⁻⁶ /°K	Yield strength, MPa
1018 steel	52	10	250
304 stainless	17	20	320

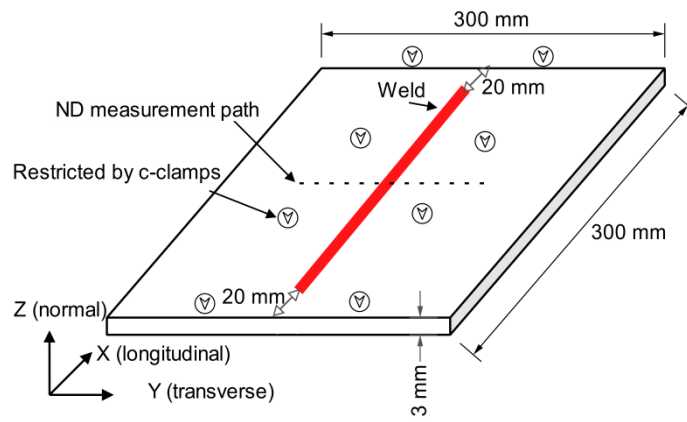


Figure 1. Dimensions and measurement directions of the specimen

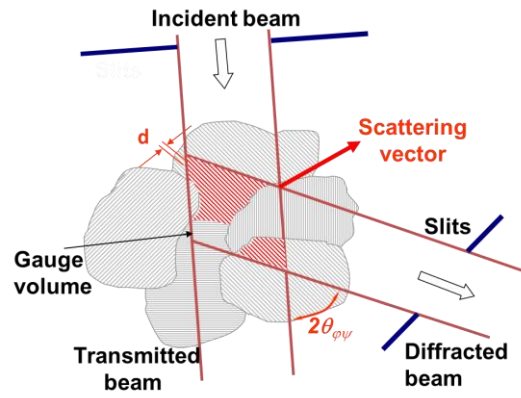


Figure 2. Scattering vector for a Bragg reflection from a single crystallographic plane family, from [26] with permission from Anna Paradowska.

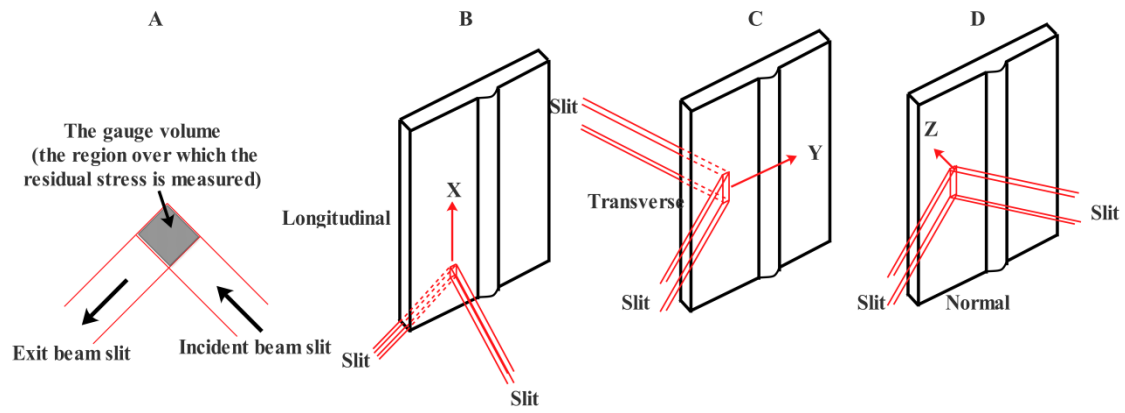


Figure 3. Gauge volume definition (a) and scattering vectors used in this study (b, c, d).

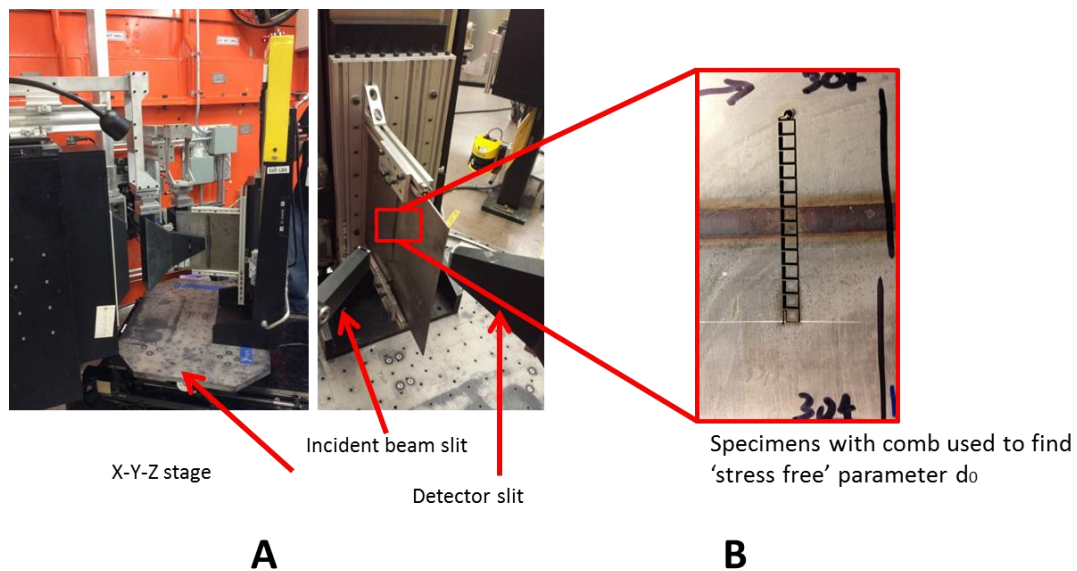


Figure 4. Experimental set up for neutron-diffraction measurements at ORNL

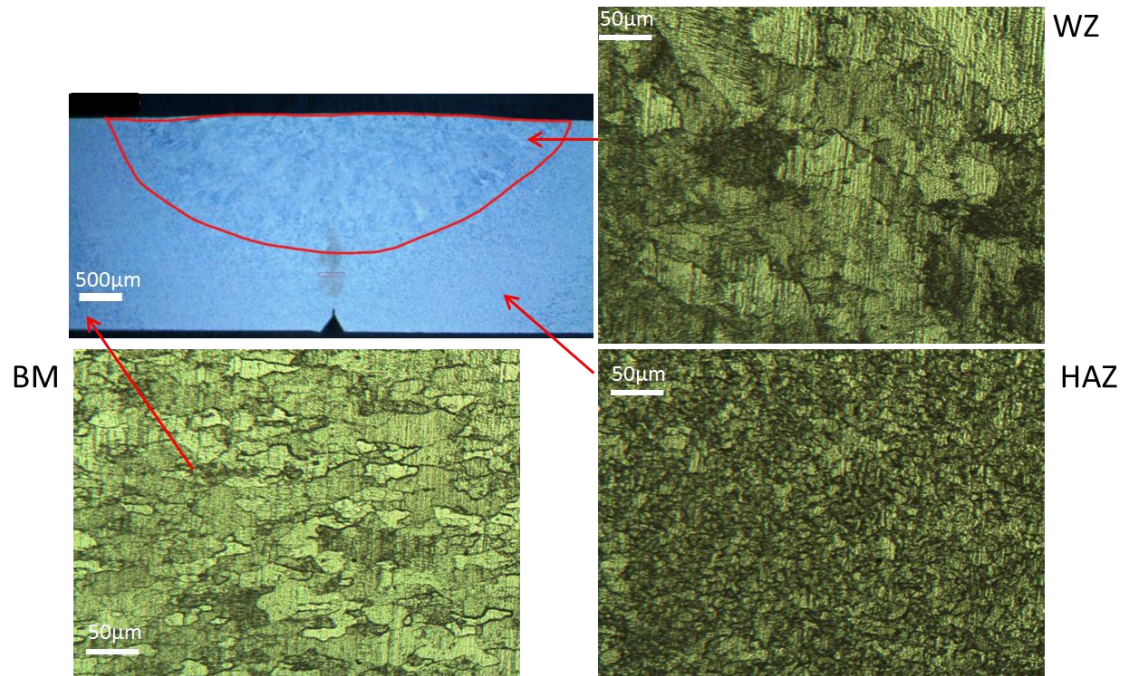


Figure 5. Macrostructure and microstructure of 1018 similar weld, etchant %2 Nital.

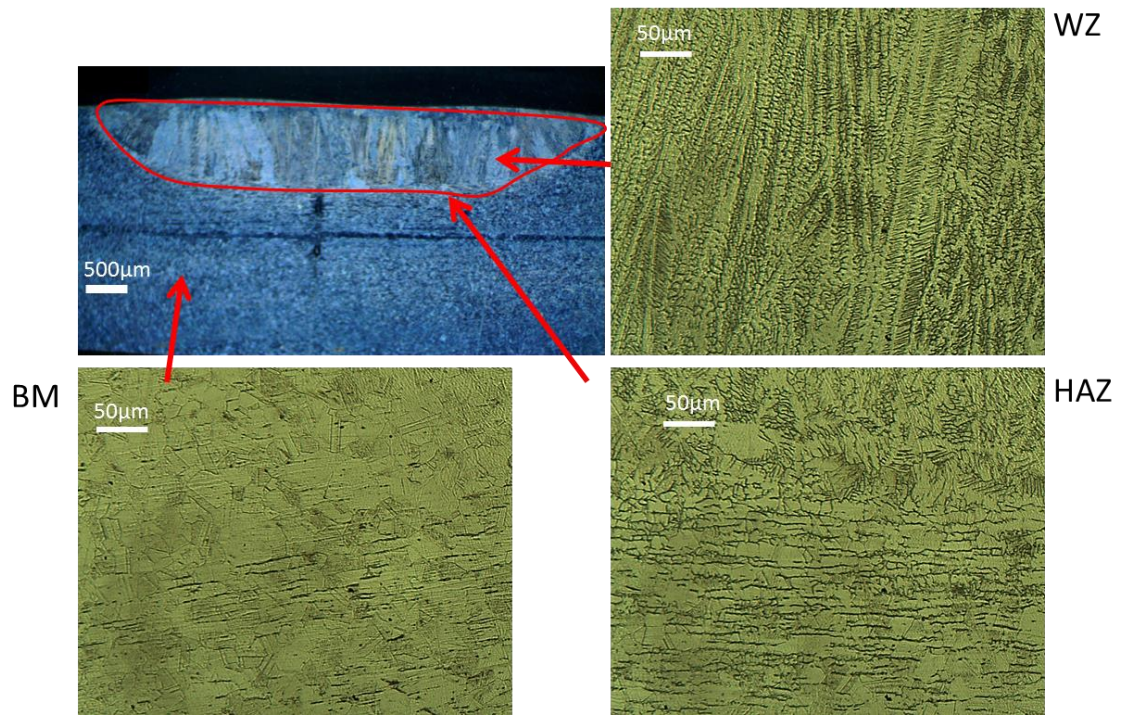


Figure 6. Macrostructure and microstructure of 304 similar weld, etchant 87% Glyceregia.

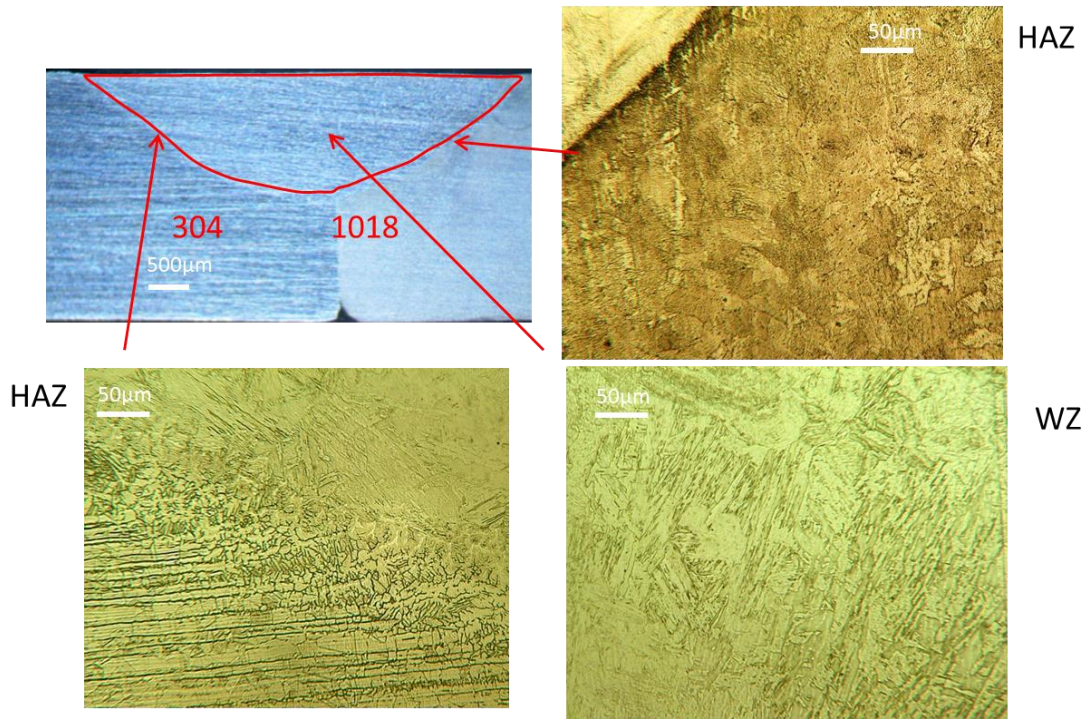


Figure 7. Macrostructure and microstructure of 304-1018 dissimilar weld, , etchant Nital %2 and 87% Glyceregia

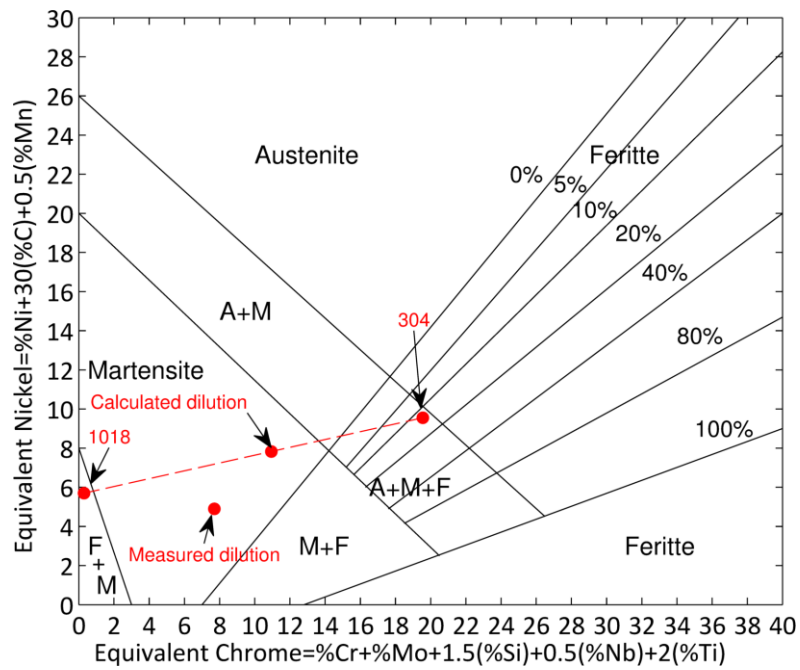


Figure 8. Estimation of weld zone phases using the Schaeffler diagram

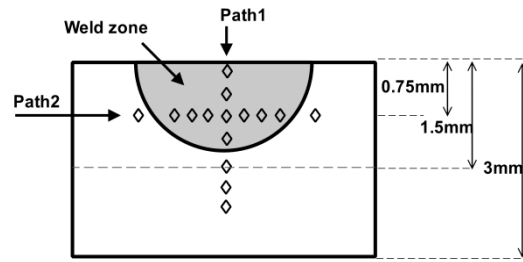


Figure 9. Microhardness measurement (paths 1&2).

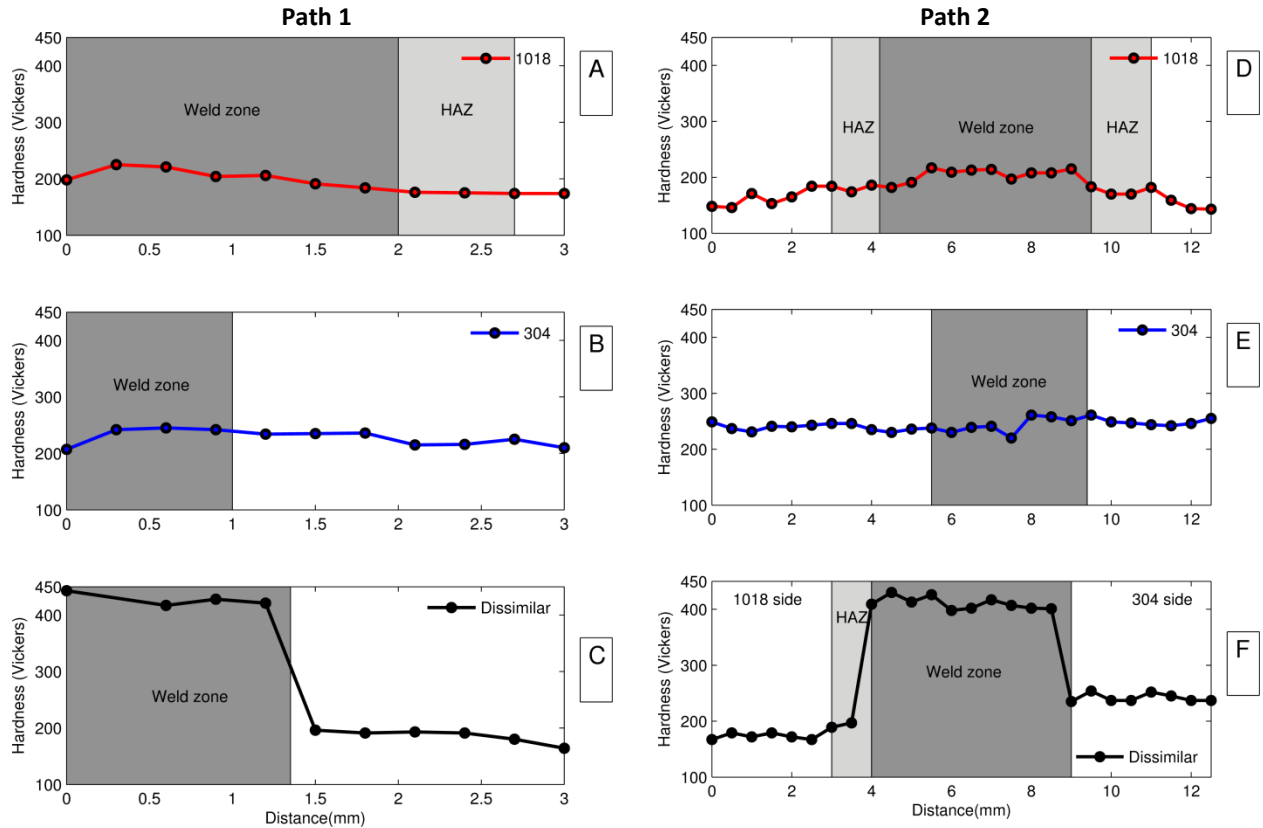


Figure 10. Micro-hardness profiles of 1018 similar, 304 similar, and 1018-304 dissimilar welds cross section.

Profile A, B, and C were taken along path 1, shown in Figure 9. Profile D, E, and F were taken along path 2, shown in Figure 9. Error of all micro-hardness measurements are $\pm 20\text{HV}$.

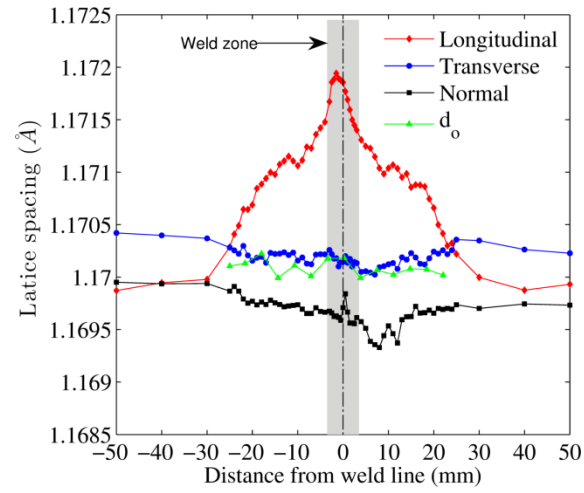


Figure 11. Distribution of the interplanar spacing (d & d_0) in 1018 similar weld. Error of lattice spacing measurement is $\pm 1 \times 10^{-4} \text{ \AA}$

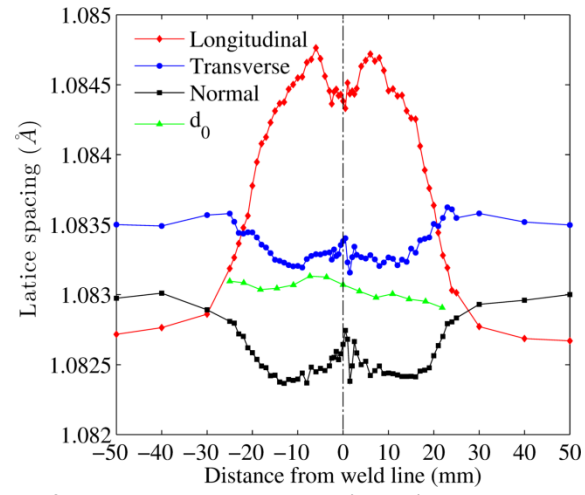


Figure 12. Distribution of the interplanar spacing (d & d_0) in 304 similar weld. Error of lattice spacing measurement is $\pm 1 \times 10^{-4} \text{ \AA}$

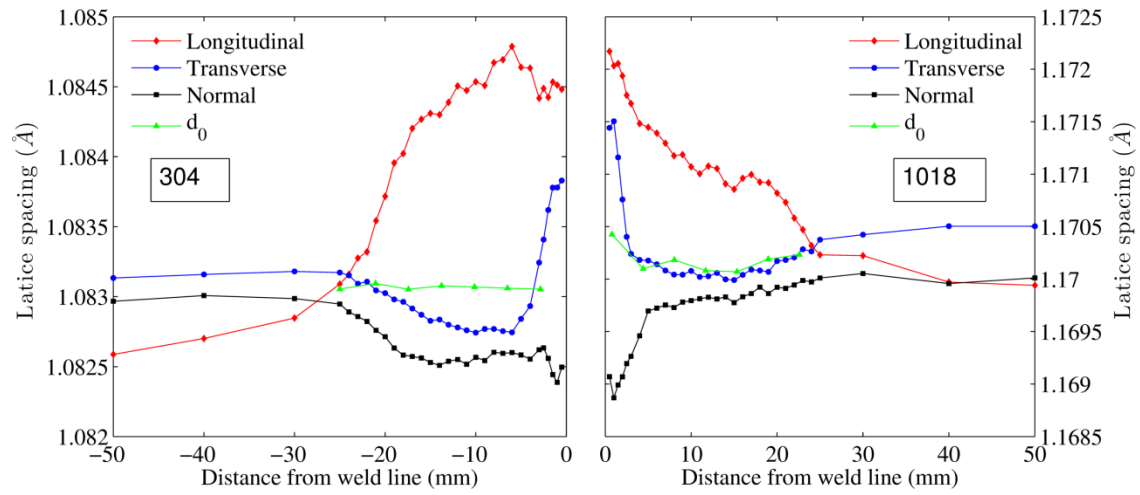


Figure 13. Distribution of the interplanar spacing (d & d_0) in 304-1018 dissimilar weld. Distance zero is the WZ. Error of lattice spacing measurement is $\pm 1 \times 10^{-4} \text{Å}$. The lattice plane measured in 304 and 1018 sides are (311) and (211), respectively.

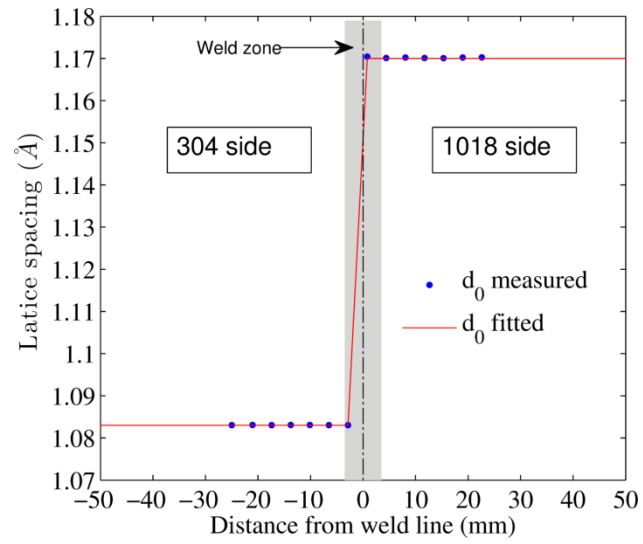


Figure 14. Distribution of the interplanar spacing (d_0) and its fitted curve in 304-1018 dissimilar weld. The fitted curve was considered for strain calculations in dissimilar weld. Distance zero is the WZ. Error of lattice spacing measurement is $\pm 1 \times 10^{-4} \text{ \AA}$

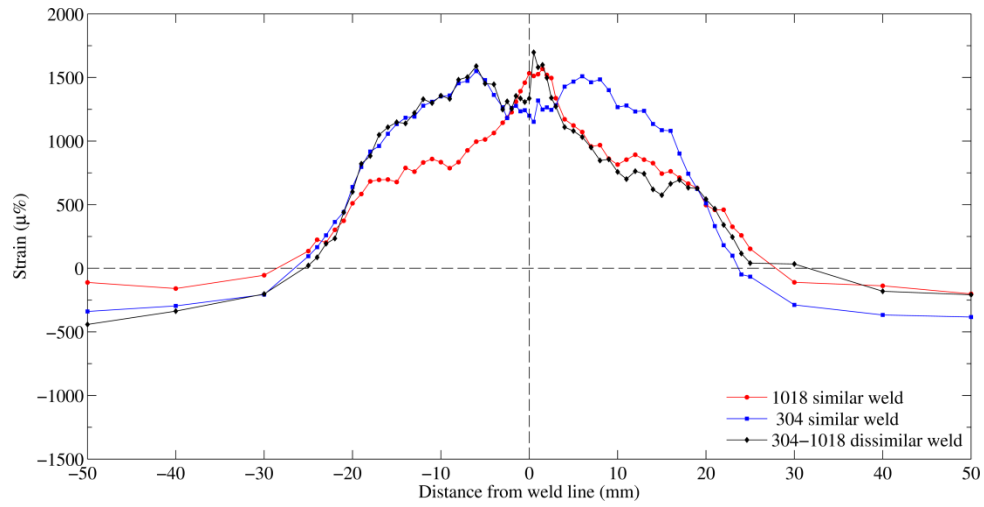


Figure 15. Distribution of longitudinal strain in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of strain measurement is $\pm 100\mu\epsilon$

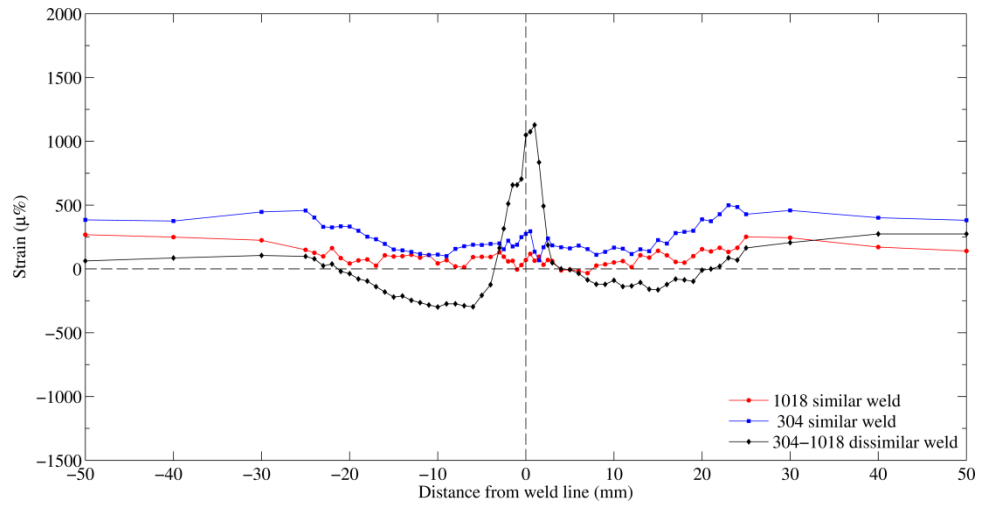


Figure 16. Distribution of transverse strain in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of strain measurement is $\pm 100\mu\epsilon$

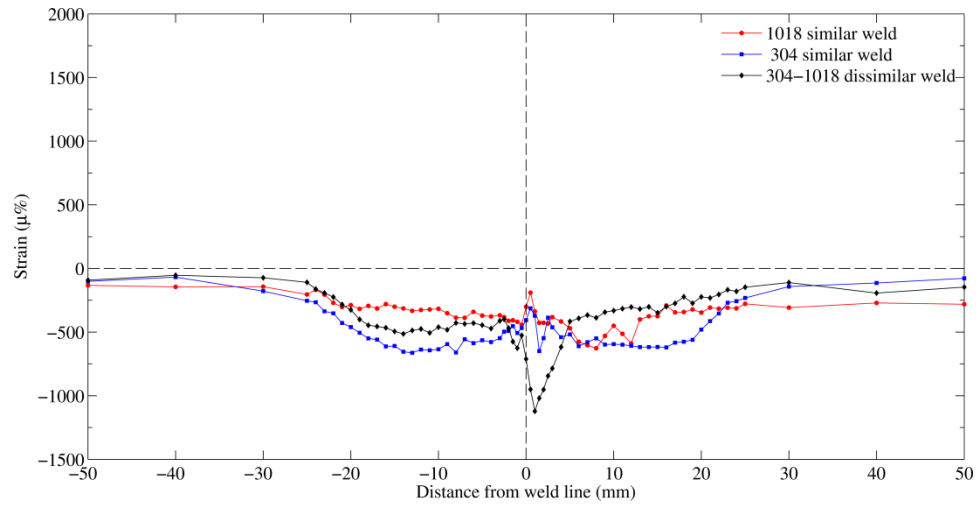


Figure 17. Distribution of normal strain in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of strain measurement is $\pm 100\mu\epsilon$.

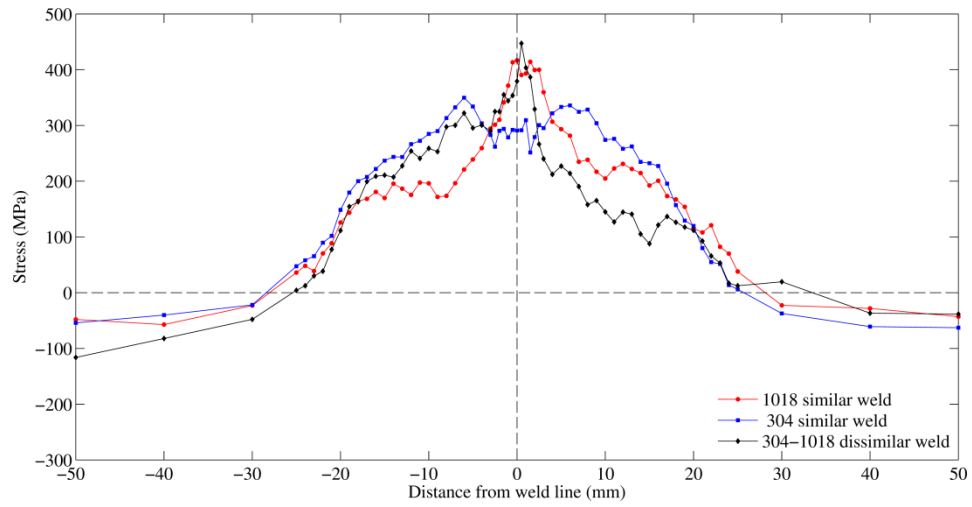


Figure 18. Distribution of longitudinal stress in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of stress measurement is ± 50 MPa.

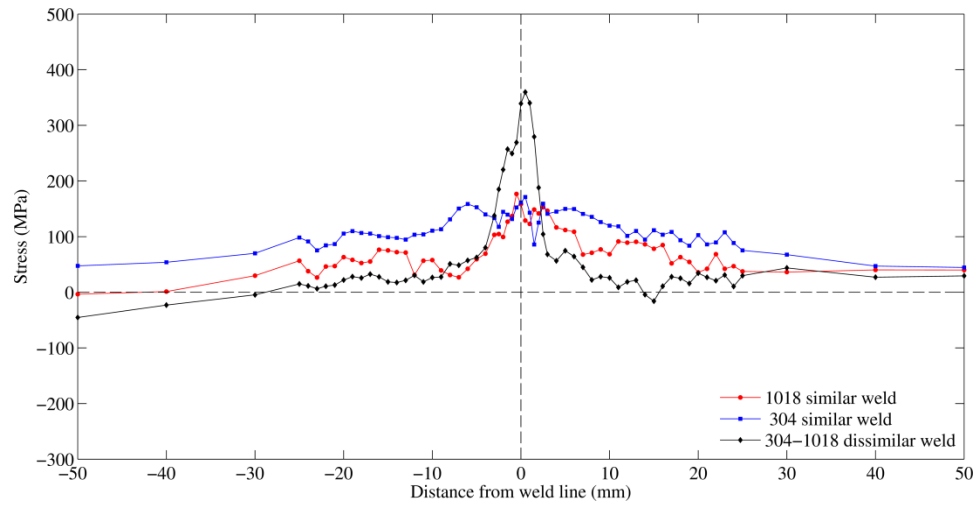


Figure 19. Distribution of transverse stress in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of stress measurement is ± 50 MPa.

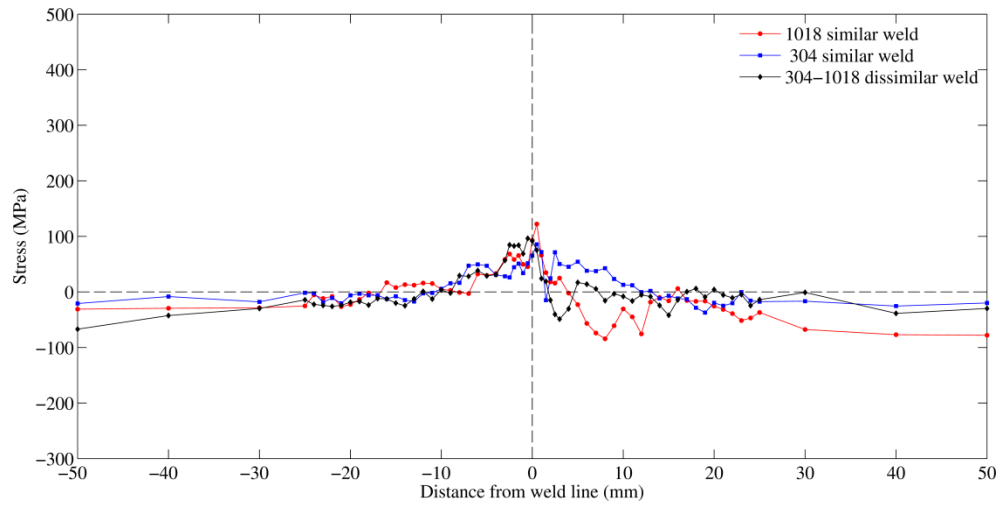


Figure 20. Distribution of normal stress in 1018 similar, 304 similar, and 304-1018 dissimilar weld. Error of stress measurement is ± 50 MPa.